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(71) Applicant: Kawasaki Steel Corporation
Kobe-shi, Hyogo-ken 651-0075 (JP)

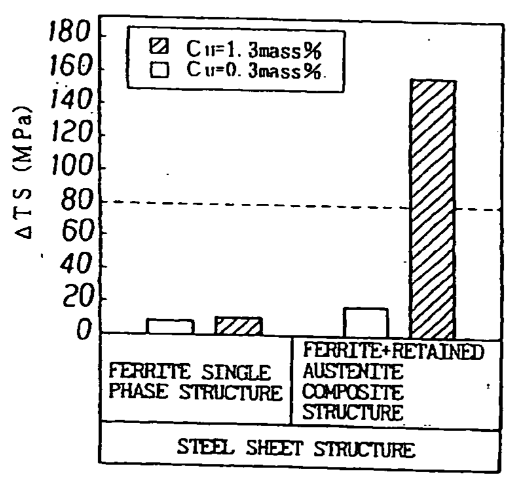
(72) Inventors:
- Matsuoka, Saiji, Kawasaki Steel Corporation
Chiba-shi, Chiba 260-0835 (JP)
- Shimizu, Tetsuo, Kawasaki Steel Corporation
Kurashiki-shi, Okayama 712-8074 (JP)
- Sakata, Kei, Kawasaki Steel Corporation
Chiba-shi, Chiba 260-0835 (JP)
- Furukimi, Osamu, Kawasaki Steel Corporation
Chiba-shi Chiba 260-0835 (JP)

(74) Representative: Grünecker, Kinkeldey,
Stockmair & Schwanhäusser Anwaltssozietät
Maximilianstrasse 58
80538 München (DE)

(54) High-ductility steel sheet excellent in press formability and strain age hardenability, and method for manufacturing the same

(57) A steel sheet composition contains appropriate amounts of C, Si, Mn, P, S, Al and N and 0.5 to 3.0% Cu. A composite structure of the steel sheet has a ferrite phase or a ferrite phase and a tempered martensite phase as a primary phase, and a secondary phase containing retained austenite in a volume ratio of not less than 1%. In place of the Cu, at least one of Mo, Cr, and W may be contained in a total amount of not more than 2.0%. This composition is useful in production of a high-ductility hot-rolled steel sheet, a high-ductility cold-rolled steel sheet and a high-ductility hot-dip galvanized steel sheet having excellent press formability and excellent stain age hardenability as represented by a ΔTS of not less than 80 MPa, in which the tensile strength increases remarkably through a heat treatment at a relatively low temperature after press forming.

Fig. 1



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manufactured by the technique disclosed in Japanese Examined Patent Application Publication No. 8-23048, the tensile strength and the yield stress increase by strain age hardening; however, a serious problem is posed in that coiling of the steel sheet at a very low coiling temperature as less than 150°C results in large variations in mechanical properties. Another problem includes a large variation in increment of yield stress after press forming and bake treatments, as well as poor press formability due to a low hole-expanding ratio (λ) and decreased stretch-flanging workability.

[0010] Japanese Unexamined Patent Application Publication No. 11-199975 proposes a hot-rolled steel sheet for working excellent in fatigue characteristics, containing C: 0.03 to 0.20%, appropriate amounts of Si, Mn, P, S and Al, Cu: 0.2 to 2.0%, and B: 0.0002 to 0.002%, of which the microstructure is a composite structure comprising ferrite as a primary phase and martensite as a second phase, and the ferrite phase contains Cu in a solid-solution and/or precipitation state of not more than 2 nm. The steel sheet disclosed in Japanese Unexamined Patent Application Publication No. 11-199975 has an object based on the fact that the fatigue limit ratio is remarkably improved only when Cu and B are added in combination, and Cu is present in an ultra fine state not more than 2 nm. For this purpose, it is essential to complete hot finish rolling at a temperature above the A_{r3} transformation point, air-cool the sheet within the temperature region of A_{r3} to A_{r1} for 1 to 10 seconds, cool the sheet at a cooling rate of not less than 20°C/second, and coil the cooled sheet at a temperature of not more than 350°C. A low coiling temperature of not more than 350°C causes serious deformation of the shape of the hot-rolled steel sheet, thus inhibiting industrially stable manufacture.

[0011] On the other hand, some automobile parts must have high corrosion resistance. A hot-dip galvanized steel sheet is suitable as a material applied to portions requiring high corrosion resistance. For this reason, a particular demand exists for hot-dip galvanized steel sheets excellent in press formability during forming, and is considerably hardened by a heat treatment after the forming.

[0012] To respond to such a demand, for example, Japanese Patent Publication No. 2802513 proposes a method for manufacturing a hot-dip galvanized steel sheet using a hot-rolled steel sheet as a black plate. The method comprises the steps of hot-rolling a steel slab containing C: not more than 0.05%, Mn: 0.05 to 0.5%, Al: not more than 0.1% and Cu: 0.8 to 2.0% at a coiling temperature of not more than 530°C. The method further comprising the subsequent steps of reducing the steel sheet surface by heating the hot-rolled steel sheet to a temperature of not more than 530°C, and hot-dip-galvanizing the sheet, whereby remarkable hardening is available through a heat treatment after forming. In the steel sheet manufactured by this method, however, the heat treatment temperature must be high as not less than 500°C, in order to obtain remarkable hardening from the heat treatment after the forming, and this has a problem in practice.

[0013] Japanese Unexamined Patent Application Publication No. 10-310824 proposes a method for manufacturing an alloyed hot-dip galvanized steel sheet having increased strength by a heat treatment after forming, using a hot-rolled or cold-rolled steel sheet as a black plate. This method comprises the steps of hot-rolling a steel containing C: 0.01 to 0.08%, appropriate amounts of Si, Mn, P, S, Al and N, and at least one of Cr, W and Mo: 0.05 to 3.0% in total. The method further comprises the step of cold-rolling or temper-rolling and annealing the sheet. The method still further comprises the step of applying hot-dip galvanizing to the sheet and heating the sheet for alloying treatment. The tensile strength of the steel sheet is increased by heating the sheet at a temperature within the range of 200 to 450°C. However, the resultant steel sheet involves a problem in that the microstructure comprises a ferrite single phase, a ferrite and pearlite composite structure, or a ferrite and bainite composite structure; hence, high ductility and low yield strength are unavailable, resulting in low press formability.

SUMMARY OF THE INVENTION

[0014] The present invention was made in view of the fact that, in spite of the strong demand as described above, a technique for industrially stably manufacturing a steel sheet satisfying these properties has never been found. The present invention solves the problems described above. It is an object of the present invention to provide is directed to high-ductility and high-strength steel sheets suitable for automobiles and having excellent press formability and excellent strain age hardenability, in which the tensile strength increases considerably through a heat treatment at a relatively low temperature after press forming. It is also an object of the present invention to provide a manufacturing method capable of stably manufacturing the high-ductility and high-strength steel sheets.

[0015] To achieve the above-mentioned object of the invention, the inventors carried out extensive studies on the effect of the steel sheet structure and alloying elements on strain age hardenability. As a result, the inventors found that a steel sheet having high age hardenability which leads to both an increase in yield stress and a remarkable increase in tensile strength can be obtained after a pre-deformation treatment with a prestrain of not less than 5% and a heat treatment at a relatively low temperature as within the range of 150 to 350°C by (1) forming a composite structure of the steel sheet comprising ferrite and a phase containing retained austenite in a volume ratio of not less than 1%, and (2) limiting the C content within the range of a low-carbon region to a medium-carbon region and containing Cu within an appropriate range or at least one of Mo, Cr, and W in place of Cu. In addition, the steel sheet was found to have satisfactory ductility, a high hole expanding ratio, and excellent press formability.

in ultra-low carbon steel or low-carbon steel in reports so far released. A reason for precipitation of very fine Cu in a heat treatment at a low temperature has not as yet been clarified to date. However, it is presumable as follows. During isothermal holding in the temperature range of 620 to 780°C or during slow cooling from this temperature range after rapid cooling subsequent to hot rolling, a large amount of Cu is distributed to the γ phase. After cooling, Cu is dissolved in the retained austenite in a supersaturation state. The retained austenite is transformed into martensite by a prestrain of not less than 5%, and very fine Cu precipitates in the strain-induced transformed martensite during a subsequent low-temperature treatment.

[0028] Next, the results of a fundamental experiment carried out by the present inventors on the cold-rolled steel sheet will be described.

[0029] A sheet bar having a composition comprising, in weight percent, C: 0.10%, Si: 1.2%, Mn: 1.4%, P: 0.01%, S: 0.005%, Al: 0.03%, N: 0.002%, and Cu: 0.3 or 1.3% was heated to 1,250°C, soaked and subjected to three-pass rolling into a thickness of 4.0 mm so that the finish rolling end temperature was 900°C. After the completion of finish rolling, a temperature holding equivalent treatment of 600°C for 1 hour was applied as a coiling treatment. Thereafter, the sheet was cold-rolled at a reduction of 70% into a cold-rolled steel sheet having a thickness of 1.2 mm. Then, the cold-rolled sheet was heated at a temperature in the range of 700 to 850°C and soaked for 60 seconds. Thereafter, the sheet was cooled to 400°C, and was held at the temperature (400°C) for 300 seconds for recrystallization annealing. By the recrystallization annealing, various cold-rolled steel sheets were obtained in which the structure changed from a single ferrite structure to a composite ferrite/retained austenite structure.

[0030] Tensile tests were conducted on the resultant cold-roll steel sheets as in the hot-rolled steel sheets to determine tensile properties. Tensile properties (YS, TS) were determined by sampling test pieces from these cold-rolled steel sheets, applying a pre-deformation treatment with a tensile prestrain of 5% to these test pieces, then heating the steel sheets at 50 to 350°C for 20 minutes, and then conducting the tensile tests.

[0031] The strain age hardenability was evaluated in terms of the tensile strength increment ΔTS from before to after the heat treatment, as in the hot-rolled steel sheet.

[0032] Fig. 3 illustrates the effect of the Cu content on the relationship between ΔTS and the recrystallization annealing temperature. The value ΔTS was determined by applying a pre-deformation treatment with a tensile prestrain of 5% to test pieces sampled from the resultant cold-rolled steel sheets, conducting a heat treatment of 250°C for 20 minutes, and carrying out a tensile test.

[0033] Fig. 3 suggests that a high strain age hardenability as represented by a ΔTS of not less than 80 MPa is available, in the case of a Cu content of 1.3 wt.%, by employing a recrystallization annealing temperature of not less than 750°C to convert the steel sheet structure into a composite ferrite/retained austenite structure. On the other hand, in the case of a Cu content of 0.3 wt.%, high strain age hardenability is unavailable because ΔTS is less than 80 MPa at any recrystallization annealing temperature. Fig. 3 suggests the possibility of manufacturing a cold-rolled steel sheet having a high strain age hardenability by optimizing the Cu content and forming a composite ferrite/retained austenite structure.

[0034] Fig. 4 illustrates the effect of the Cu content on the relationship between ΔTS and the heat treatment temperature after pre-strain treatment. The steel sheet used was annealed at 800°C, which was the dual phase region of ferrite (α) + austenite (γ), for a holding time of 60 seconds after cold rolling, cooled from the holding temperature (800°C) to 400°C at a cooling rate of 30°C/second, and held at 400°C for 300 seconds. The steel sheets had a composite ferrite/retained austenite (secondary phase) microstructure, the volume ratio of the retained austenite structure being 4%.

[0035] Fig. 4 shows that the increment ΔTS increases as the heat treatment temperature increases and strongly depends on the Cu content. With a Cu content of 1.3 wt.%, a high strain age hardenability as represented by a ΔTS of not less than 80 MPa is obtained at a heat treatment temperature of not less than 150°C. For a Cu content of 0.3 wt.%, ΔTS is less than 80 MPa at any heat treatment temperature, and high strain age hardenability cannot be obtained.

[0036] In addition, a hole expanding test was carried on cold-rolled steel sheets having a composite ferrite/retained austenite structure and Cu contents of 0.3 wt% and 1.3 wt.% to determine the hole expanding ratio (λ), as in the hot-rolled steel sheet.

[0037] In the cold-rolled steel sheet with a Cu content of 1.3%, λ was 130%; while in the cold-rolled steel sheet with a Cu content of 0.3 %, λ was 60%. It is clear that, for a Cu content of 1.3 wt.%, the hole expanding ratio is increased and hole expanding formability is improved even in the cold-rolled steel sheet, as in the hot-rolled steel sheet. A detailed mechanism of improvement in hole expanding formability with content of Cu has not yet been clarified, as in the hot-rolled steel sheet. Also, in the cold-rolled steel sheet, it is considered that the contained Cu reduces the difference in hardness between the ferrite/retained austenite structure and the strain-induced transformed martensite structure.

[0038] In the cold-rolled steel sheet of the present invention, very fine Cu precipitates in the steel sheet as a result of a pre-deformation with a strain larger than 2%, which is equivalent to the prestrain on measuring the deformation stress increment from before to after a usual heat treatment, and a heat treatment at a relatively low temperature of 150 to 350°C. According to a study carried out by the present inventors, also in the cold-rolled steel sheet, high strain

structure of ferrite/tempered martensite/retained austenite and Cu contents of 0.3 wt% and 1.3 wt.% to determine the hole expanding ratio (λ), as in the hot-rolled steel sheet and the cold-rolled steel sheet.

[0049] The hole expanding ratio λ of the steel sheet having a Cu content of 1.3% was 120%, while the hole expanding ratio λ of the steel sheet having a Cu content of 0.3% was 50%. The results suggest that for a Cu content of 1.3 wt%, the hole expanding ratio is increased and hole expanding formability is improved, as compared with a Cu content of 0.3%.

[0050] A detailed mechanism of improvement in hole expanding formability with content of Cu has not yet been clarified, as in the hot-rolled steel sheet and the cold-rolled steel sheet, but it is considered that the contained Cu reduces the difference in hardness among the ferrite, the tempered martensite/retained austenite, and the martensite formed by strain induced transformation.

[0051] On the basis of the novel findings as described above, the present inventors carried out further extensive studies and found that the above-mentioned phenomena occurred in steel sheets not containing Cu as well.

[0052] The structure of a steel sheet having a composition containing at least one of Mo, Cr, and W was converted to a composite structure containing a ferrite primary phase and a phase containing retained austenite as a secondary phase. Thereafter, by applying a prestrain and a heat treatment in a low temperature region, it was found that very fine carbides precipitated in the strain-induced transformed martensite, resulting in an increase in tensile strength. The strain-induced fine precipitation at a low temperature was more remarkable in a steel composition containing at least one of Nb, Ti, and V in addition to at least one of Mo, Cr, and W.

[0053] The present invention was completed through further studies on the basis of the aforementioned findings. The gist of the present invention is as follows:

(1) A high-ductility steel sheet excellent in press formability and in strain age hardenability as represented by a ΔTS of not less than 80 MPa, comprising a composite structure containing a primary phase containing a ferrite phase and a secondary phase containing a retained austenite phase in a volume ratio of not less than 1%.
(2) A high-ductility steel sheet according to aspect (1), wherein the steel sheet is a hot-rolled steel sheet, and the primary phase consisting essentially of a ferrite phase.
(3) A high-ductility steel sheet according to aspect (2), wherein the hot-rolled steel sheet has a composition comprising, in weight percent, C: 0.05 to 0.20%, Si: 1.0 to 3.0%, Mn: not more than 3.0%, P: not more than 0.10%, S: not more than 0.02%, Al: not more than 0.30%, N: not more than 0.02%, and Cu: 0.5 to 3.0%, and the balance Fe and incidental impurities.
(4) A high-ductility steel sheet according to aspect (3), the composition further comprising, in weight percent, at least one of the following Groups A to C:

Group A: Ni: not more than 2.0%;
Group B: at least one of Cr and Mo: not more than 2.0% in total; and
Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

(5) A high-ductility steel sheet according to aspect (2), wherein the hot-rolled steel sheet has a composition comprising, in weight percent, C: 0.05 to 0.20%, Si: 1.0 to 3.0%, Mn: not more than 3.0%, P: not more than 0.10%, S: not more than 0.02%, Al: not more than 0.30%, N: not more than 0.02%, at least one of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0%, not more than 2.0% in total, and the balance Fe and incidental impurities.
(6) A high-ductility steel sheet according to aspect (5), the composition further containing, in weight percent, at least one of Nb, Ti, and V in an amount of not more than 2.0% in total.
(7) A method for manufacturing a high-ductility hot-rolled steel sheet excellent in press formability and in strain age hardenability as represented by a ΔTS of not less than 80 MPa, comprising the steps of: hot-rolling a steel slab having a composition comprising, in weight percent, C: not more than 0.20%, Si: 1.0 to 3.0%, Mn: not more than 3.0%, P: not more than 0.10%, S: not more than 0.02%, Al: not more than 0.30%, N: not more than 0.02%, and Cu: 0.5 to 3.0%, into a hot-rolled steel sheet having a prescribed thickness, the hot rolling step including finish-rolling at a finish rolling end temperature of 780 to 980°C; cooling the finish-rolled steel sheet to a temperature in the range of 620 to 780°C within 2 seconds at a cooling rate of at least 50°C/second; holding the sheet at the temperature in the range of 620 to 780°C for 1 to 10 seconds, or slowly cooling the sheet at a cooling rate of not more than 20°C/second; cooling the sheet at a cooling rate of not less than 50°C/second to a temperature of 300 to 500°C; and coiling the sheet.
(8) A method for manufacturing a high-ductility hot-rolled steel sheet excellent in press formability and in strain age hardenability as represented by a ΔTS of at least 80 MPa, according to aspect (7), the composition further comprising, in weight percent, at least one of the following Groups A to C:

Group A: Ni: not more than 2.0%;

the slab at a finish rolling end temperature of not less than 700°C, and coiling the hot-rolled steel sheet at a coiling temperature of not more than 800°C.
(22) A method for manufacturing a cold-rolled steel sheet according to any one of aspects (17) to (21), wherein all or part of the hot rolling is lubrication rolling.
(23) A high-ductility hot-dip galvanized steel sheet comprising a hot-dip galvanizing layer or an alloyed hot-dip galvanizing layer formed on the surface of the high-ductility steel sheet according to any one of aspects (1) to (6).
(24) A high-ductility hot-dip galvanized steel sheet comprising a hot-dip galvanizing layer or an alloyed hot-dip galvanizing layer formed on the surface of the high-ductility steel sheet according to any one of aspects (12) to (16).
(25) A high-ductility steel sheet according to aspect (1), wherein the steel sheet is a hot-dip galvanized steel sheet having a hot-dip galvanizing layer or an alloyed hot-dip galvanizing layer formed on a surface of the steel sheet, and the primary phase containing a ferrite phase comprises a ferrite phase and a tempered martensite phase.
(26) A high-ductility steel sheet according to aspect (25), wherein the steel sheet has a composition comprising, in weight percent, C: not more than 0.20%, Si: not more than 2.0%, Mn: not more than 3.0%, P: not more than 0.1%, S: not more than 0.02%, Al: not more than 0.3%, N: not more than 0.02%, Cu: 0.5 to 3.0%, and the balance Fe and incidental impurities.
(27) A high-ductility steel sheet according to aspect (26), the composition further containing, in weight percent, at least one of the following Groups A to C:

Group A: Ni: not more than 2.0%;
Group B: at least one of Cr and Mo: not more than 2.0% in total; and
Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

(28) A high-ductility steel sheet according to aspect (25), wherein the steel sheet has a composition comprising, in weight percent, C: not more than 0.20%, Si: not more than 2.0%, Mn: not more than 3.0%, P: not more than 0.1%, S: not more than 0.02%, Al: not more than 0.3%, N: not more than 0.02%, at least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0% in a total amount of not more than 2.0%, and the balance Fe and incidental impurities.
(29) A high-ductility steel sheet according to aspect (28), the composition further containing, in weight percent, at least one of Nb, Ti, and V in a total amount of not more than 2.0%.
(30) A method for manufacturing of a high-ductility hot-dip galvanized steel sheet excellent in press formability and in strain age hardenability as represented by a ΔTS of not less than 80 MPa, comprising: a primary heat-treating step of heating a steel sheet to a temperature of not less than the A_{C1} transformation point and rapidly cooling the steel sheet, the steel sheet having a composition containing, in weight percent, C: not more than 0.20%, Si: not more than 2.0%, Mn: not more than 3.0%, P: not more than 0.1%, S: not more than 0.02%, Al: not more than 0.3%, N: not more than 0.02%, and Cu: 0.5 to 3.0%; a secondary heat-treating step of heating the steel sheet to a temperature in the range of the A_{C1} transformation point to the A_{C3} transformation point; and a hot-dip galvanizing step of forming a hot-dip galvanizing layer on the surface of the steel sheet.
(31) A method for manufacturing a high-ductility cold-rolled steel sheet according to aspect (30), the composition further containing, in weight percent, at least one of the following Groups A to C:

Group A: Ni: not more than 2.0%;
Group B: at least one of Cr and Mo: not more than 2.0% in total; and
Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

(32) A method for manufacturing a high-ductility hot-dip galvanized steel according to aspect (30), wherein the steel sheet is replaced with a steel sheet having a composition comprising, in weight percent, C: not more than 0.20%, Si: not more than 2.0%, Mn: not more than 3.0%, P: not more than 0.1%, S: not more than 0.02%, Al: not more than 0.3%, N: not more than 0.02%, and at least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0% in a total amount of not more than 2.0%.
(33) A method for manufacturing a high-ductility hot-dip galvanized steel sheet according to aspect (32), the composition further containing, in weight percent, at least one of Nb, Ti, and V in a total amount of not more than 2.0%.
(34) A method for manufacturing a high-ductility hot-dip galvanized steel sheet according to any one of aspects (30) to (33), further comprising a pickling treatment step of pickling the steel sheet between the primary heat-treating step and the secondary heat-treating step.
(35) A method for manufacturing a high-ductility hot-dip galvanized steel sheet according to any one of aspects (30) to (34), further comprising an alloying step of alloying the hot-dip galvanizing layer, subsequent to the hot-dip galvanizing step.
(36) A method for manufacturing a high-strength hot-dip galvanized steel sheet according to any one of aspects

[0061] Conventional bake treatment conditions include 170°C × 20 minutes as a standard. If precipitation strengthening by very fine Cu or fine carbide is performed as in the present invention, the heat treatment temperature must be 150°C or more. Under conditions including a temperature exceeding 350°C, on the other hand, the strengthening effect is saturated, and the steel sheet tends to soften. Heating to a temperature exceeding 350°C causes marked occurrence of thermal strain or temper color. For these reasons, a heat treatment temperature in the range of 150 to 350°C is adopted for strain age hardening in the present invention. The holding time of the heat treatment temperature should be at least 30 seconds. Holding a heat treatment temperature in the range of 150 to 350°C for about 30 seconds permits achievement of substantially satisfactory strain age hardening. For further enhanced strain age hardening, the holding time is preferably at least 60 seconds, and more preferably at least 300 seconds.

[0062] The heat treatment method after the pre-deformation is not limited in the present invention, and atmospheric heating in a furnace in general bake treatment, induction heating, non-oxidizing flame heating, laser heating, and plasma heating are suitably applicable. So-called hot pressing for pressing a heated steel sheet is also very effective means in the present invention.

[0063] Next, the hot-rolled steel sheet, the cold-rolled steel sheet, and the hot-dip galvanized steel sheet in the present invention will be described individually.

(1) Hot-rolled steel sheet

[0064] The hot-rolled steel sheet of the present invention will now be described.

[0065] The hot-rolled steel sheet of the present invention has a composite structure comprising a ferrite primary phase and a secondary phase containing a retained austenite phase having a volume ratio of not less than 1% of the entire structure. As described above, a hot-rolled steel sheet having such a composite structure exhibits high ductility, high strength-ductility balance (TS × El), and excellent press formability.

[0066] Ferrite primary phase is preferably present in a volume ratio of not less than 50%. With a ferrite phase of less than 50%, it is difficult to keep high ductility, resulting in lower press formability. When further enhanced ductility is required, the volume ratio of the ferrite phase is preferably not less than 80%. For the purpose of making full use of advantages of the composite structure, the ferrite phase is preferably not more than 98%.

[0067] In the present invention, steel must contain retained austenite phase as the secondary phase in a volume ratio of not less than 1% of the entire structure. With a retained austenite phase of less than 1%, high elongation (El) cannot be obtained. To obtain higher elongation (El), the retained austenite phase content is preferably not less than 2% and more preferably not less than 3%.

[0068] The secondary phase may be a single retained austenite phase having a volume ratio of not less than 1%, or may be a mixture of a retained austenite phase of a volume ratio of not less than 1% and another phase, i.e., a pearlite phase, a bainite phase, and/or a martensite phase.

[0069] The reasons for limiting the composition of the hot-rolled steel sheet of the present invention will now be described. The weight percent in the composition will hereafter be denoted simply as %.

C: 0.05 to 0.20%

[0070] C is an element, which improves strength of a steel sheet and promotes the formation of a composite structure of ferrite and retained austenite, and is preferably contained in an amount of not less than 0.05% for forming the composite structure according to the present invention. A C content exceeding 0.20% causes an increase in proportions of carbides in steel, resulting in a decrease in ductility, and hence a decrease in press formability. A more serious problem is that a C content exceeding 0.20% leads to significant deterioration of spot weldability and arc weldability. For these reasons, the C content is limited within the range of 0.05 to 0.20% in the present invention. From the viewpoint of formability, the C content is preferably not more than 0.18%.

Si: 1.0 to 3.0%

[0071] Si is a useful strengthening element, which improves the strength of a steel sheet without a marked decrease in ductility of the steel sheet. In addition, Si is necessary for forming a retained austenite phase. To obtain these effects, Si is preferably contained in an amount of not less than 1.0% and more preferably not less than 1.2%. An Si content exceeding 3.0% leads to deterioration of press formability and degrades the surface quality. The Si content is therefore limited within the range of 1.0 to 3.0%.

Mn: not more than 3.0%

[0072] Mn is a useful element, which strengthens steel and prevents hot cracking caused by S, and is therefore

Group A: Ni: not more than 2.0%

[0079] Group A: Ni is effective for preventing the formation of surface defects on the steel sheet surface containing Cu, and may be added as required. The Ni content is preferably about a half the Cu content, i.e., in the range of about 30 to about 80% of the Cu content. An Ni content exceeding 2.0% cannot give further enhancement in the effect because saturation of the effect, leading to economic disadvantages, and causes deterioration of press formability. For these reasons, the Ni content is preferably limited to not more than 2.0%.

Group B: at least one of Cr and Mo: not more than 2.0% in total

[0080] Group B: Both Cr and Mo, as well as Mn, strengthen the steel sheet and at least one thereof can be contained as required. This effect is particularly remarkable at a Cr content of not less than 0.1% and at an Mo content of not less than 0.1%. It is therefore preferable to contain at least one of Cr: not less than 0.1% and Mo: not less than 0.1%. If at least one of Cr and Mo are contained in a total amount exceeding 2.0%, press formability is impaired. It is therefore preferable to limit the total content of Cr and Mo to not more than 2.0%.

Group C: at least one of Nb, Ti, and V: not more than 0.2% in total

[0081] Group C: Nb, Ti, and V are carbide-forming elements and effectively increase the strength by fine dispersion of carbides, and can be selected and contained as required. This effect can be achieved at an Nb content of not less than 0.01%, a Ti content of not less than 0.01%, and a V content of not less than 0.01%. However, a total content of Nb, Ti, and V exceeding 0.2% causes deterioration of press formability. Thus, the total content of Nb, Ti, and V is preferably limited to not more than 0.2%.

[0082] In the present invention, in place of the aforementioned Cu or at least one of the above-mentioned Groups A to C, at least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0%, and W: 0.05 to 2.0% may be contained in an amount of not more than 2.0% in total, and at least one selected from the group consisting of Nb, Ti, and V may be further contained in an amount of not more than 2.0% in total.

At least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0%, in an amount of not more than 2.0% in total

[0083] Mo, Cr, and W are elements, which remarkably increase strain age hardening (increase in strength after pre-deformation and heat treatment) of a steel sheet, and are one of the most important elements in the present invention. That is, in the present invention, a hot-rolled steel sheet having a composite structure containing ferrite as a primary phase and a secondary phase of retained austenite and containing at least one of Mo, Cr, and W, causes strain-induced transformation of the retained austenite into martensite when a prestrain of not less than 5% and a low-temperature heat treatment are applied to the hot-rolled steel sheet, and strain-induced fine precipitation of fine carbides at a low temperature occurs in the strain-induced transformed martensite, resulting in an increase in tensile strength ΔTS of not less than 80 MPa. With a content of at least one of Mo, Cr, and W of less than 0.05%, changing the steel sheet structure and pre-deformation and heat treatment conditions does not cause an increase in tensile strength ΔTS of not less than 80 MPa. On the other hand, a content of at least one of Mo, Cr, and W exceeding 2.0% does not give a corresponding effect because of saturation of the effect, leading to economic disadvantages, and causes deterioration of press formability. The contents of Mo, Cr, and W are each preferably limited within the range of 0.05 to 2.0%. From the viewpoint of press formability, the total content of Mo, Cr and/or W is more preferably limited to not more than 2.0%.

At least one of Nb, Ti, and V, in a total amount of not more than 2.0%

[0084] Nb, Ti, and V are carbide-forming elements, and can be added as required. Containing at least one of Nb, Ti, and V, in addition to at least one of Mo, Cr, and W, and forming a composite structure containing a ferrite primary phase and a secondary phase of retained austenite form fine carbides in the strain-induced transformed martensite and cause strain-induced precipitation at low temperature, resulting in an increase in tensile strength ΔTS of not less than 80 MPa. In order to obtain these effects, an Nb content is preferably not less than 0.01%, a Ti content is preferably not less than 0.01%, and a V content is preferably not less than 0.01%, and at least one of Nb, Ti, and V can be added as required. However, a total content exceeding 2.0% causes deterioration of press formability. Thus, the total content of Nb, Ti, and V is preferably limited to not more than 2.0%.

[0085] Apart from the above-mentioned elements, at least one of Ca: not less than 0.1% and REM: not less than 0.1% may be contained. Ca and REM are elements contributing to improvement in stretch flanging property through conformational control of inclusions. If the Ca content exceeds 0.1% or the REM content exceeds 0.1%, however, there

second causes insufficient concentration of carbon into the austenite. On the other hand, a time exceeding 10 seconds causes pearlite transformation.

[0102] A cooling rate of the slow cooling treatment exceeding 20°C/second causes insufficient concentration of carbon into the austenite.

[0103] After the isothermal holding treatment or slow cooling treatment, the rolled sheet is preferably cooled again to a temperature of 300 to 500°C at a cooling rate of not less than 50°C/second, and then coiled. That is, the rolled sheet is preferably coiled at a coiling temperature (CT) of 300 to 500°C.

[0104] After the isothermal holding treatment or slow cooling treatment, the rolled sheet is cooled to a temperature of 300 to 500°C. Also, the cooling rate of this treatment is preferably not less than 50°C/second. With the cooling rate of less than 50°C/second, pearlite transformation occurs and ductility is decreased. The cooling rate is more preferably within the range of 50 to 200°C/second.

[0105] With a coiling temperature CT of less than 300°C, the secondary phase contains martensite. On the other hand, with the coiling temperature exceeding 500°C, the secondary phase contains pearlite. Thus, the coiling temperature CT is preferably within a range of 300 to 500°C.

[0106] In the present invention, all or part of finish rolling may be lubrication rolling to reduce the rolling load during hot rolling. Application of lubrication rolling is effective also from the viewpoint of achieving a uniform steel sheet shape and uniform material quality. The frictional coefficient on the lubrication rolling is preferably in the range of 0.25 to 0.10. A continuous rolling process is preferable one,in which neighboring sheet bars can be connected to each other to perform finish rolling continuously. Application of the continuous rolling process is desirable also from the viewpoint of operational stability of hot rolling.

[0107] After the completion of hot rolling, temper rolling of not more than 10% may be applied for adjustment such as shape correction or surface roughness control.

[0108] The hot-rolled steel sheet of the invention may be used as a steel sheet for processing and as a steel sheet for surface treatments. Surface treatments include galvanizing (including alloying), tin-plating and enameling. After annealing or galvanizing, the hot-rolled steel sheet of the present invention may be subjected to a special treatment to improve activity to chemical treatment, weldability, press formability, and corrosion resistance.

(2) Cold-rolled steel sheet

[0109] A cold-rolled steel sheet of the present invention will now be described.

[0110] The cold-rolled steel sheet of the present invention has a composite structure comprising a ferrite primary phase and a secondary phase containing retained austenite having a volume ratio of not less than 1% of the entire structure. As described above, a cold-rolled steel sheet having such a composite structure exhibits high elongation (El), high strength/elongation balance (TS × El), and excellent press formability.

[0111] The volume ratio of the ferrite primary phase contained in the composite structure is preferably not less than 50%. With a ferrite phase content of less than 50%, it is difficult to keep high ductility, resulting in poor press formability. When further enhanced ductility is required, the volume ratio of the ferrite phase is preferably not less than 80%. For the purpose of making full use of advantages of the composite structure, the ferrite phase is preferably not more than 98%.

[0112] In the present invention, the steel sheet must contain a retained austenite phase as the secondary phase in a volume ratio of not less than 1% of the entire structure. With a retained austenite phase content of less than 1%, it is impossible to obtain high elongation (El). To obtain higher elongation (El), the retained austenite phase is preferably contained in a volume ratio of not less than 2%, more preferably, not less than 3%.

[0113] The secondary phase may be a single retained austenite phase having a volume ratio of not less than 1%, or may be a mixture of a retained austenite phase of a volume ratio of not less than 1% and an auxiliary (another) phase comprising a pearlite phase, a bainite phase, and/or a martensite phase.

[0114] The reasons for limiting the composition of the cold-rolled steel sheet of the present invention will now be described. The weight percent in the composition will simply be denoted hereinafter as %.

C: not more than 0.20%

[0115] C is an element, which improves strength of a steel sheet and promotes the formation of a composite structure of a ferrite phase and a retained austenite phase, and is preferably contained in an amount of not less than 0.01% from the viewpoint of forming the retained austenite phase in the present invention. A C content is more preferably not less than 0.05%. A C content exceeding 0.20%, however, causes an increase in amount of carbides in the steel, resulting in a decrease in ductility, and hence a decrease in press formability. A more serious problem is that a C content exceeding 0.20% leads to remarkable deterioration of spot weldability and arc weldability. For these reasons, in the present invention, the C content is limited to not more than 0.20%. From the viewpoint of formability, the C content is

effects. Furthermore, deterioration of press formability occurs, and the surface quality of the steel sheet is degraded. The Cu content is, therefore, limited within the range of 0.5 to 3.0%. In order to simultaneously achieve a higher ΔTS and excellent press formability, the Cu content is preferably within the range of 1.0 to 2.5%.

[0123] In the present invention, the above-mentioned composition containing Cu preferably further contains, in weight percent, at least one of the following Groups A to C:

Group A: Ni: not more than 2.0%;
Group B: at least one of Cr and Mo: not more than 2.0% in total; and
Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

Group A: Ni: not more than 2.0%

[0124] Group A: Ni is an element effective for preventing surface defects produced by Cu contained in the steel sheet, and may be contained as required. The Ni content depends on the Cu content, and is preferably about a half the Cu content, more specifically, within the range of about 30 to about 80% of the Cu content. An Ni content exceeding 2.0% cannot give further enhancement in the effect because of saturation of the effect, leading to economic disadvantages, and causes deterioration of press formability. For these reasons, the Ni content is preferably limited to not more than 2.0%.

Group B: at least one of Cr and Mo: not more than 2.0% in total

[0125] Group B: Both Cr and Mo, as well as Mn, strengthen the steel sheet and may be contained as required preferably in an amount of not less than 0.1% for Cr and not less than 0.1% for Mo. If at least one of Cr and Mo are contained in an amount exceeding 2.0% in total, press formability is impaired. It is therefore preferable to limit the total content of Cr and Mo forming Group B to not more than 2.0%.

Group C: at least one of Nb, Ti, and V: not more than 0.2% in total

[0126] Group C: Nb, Ti, and V are elements, which effectively form fine dispersion of carbides contributing to an increase in strength. Therefore, Nb, Ti, and V can be selected and contained as required preferably in an amount of not less than 0.01% for Nb, in an amount of not less than 0.01% for Ti and in an amount of not less than 0.01% for V. If the total content of at least one of Nb, Ti, and V exceeds 0.2%, the press formability is impaired. Thus, the total content of Nb, Ti and/or V is preferably limited to not more than 0.2%.

[0127] In the present invention, in place of the aforementioned Cu, at least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0%, and W: 0.05 to 2.0% may be contained in an amount of not more than 2.0% in total.

At least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0%, in an amount of not more than 2.0% in total

[0128] In the present invention, all of Mo, Cr, and W, as well as Cu, are the most important elements, which remarkably increase strain age hardening of the steel sheet, and can be selected and contained. When a steel sheet containing at least one of Mo, Cr, and W and having a composite structure of a ferrite phase and a phase containing retained austenite is subjected to a prestrain (pre-deformation) of not less than 5% and a low-temperature heat treatment (heat treatment), the retained austenite is changed into martensite by strain-induced transformation. Then, the formation of fine carbide precipitation in the martensite is induced by the strain, resulting in an increase in tensile strength ΔTS of not less than 80 MPa. With a content of each of these elements of less than 0.05%, changing pre-deformation/heat treatment conditions does not give an increase in tensile strength ΔTS of at least 80 MPa. If the content of each of these elements exceeds 2.0%, a further enhanced effect corresponding to the content cannot be expected as a result of saturation of the effect, leading to economic disadvantages, and this results in deterioration of press formability. The contents of Mo, Cr, and W are therefore limited within the range of 0.05 to 2.0% for Mo, 0.05 to 2.0% for Cr, and 0.05 to 2.0% for W. From the viewpoint of press formability, the total content of Mo, Cr, and W is limited to not more than 2.0%.

[0129] In the present invention, at least one selected from the group consisting of Mo, Cr, and W is preferably contained and further, at least one of Nb, Ti, and V are preferably contained not more than 2.0% in total.

[0130] At least one of Nb, Ti, and V, in a total amount of not more than 2.0%:

Nb, Ti, and V are elements forming carbides, and can be selected and contained as required, when at least one of Mo, Cr, and W is added. When the steel composition contains at least one of Mo, Cr, and W and has a composite structure containing a ferrite phase and a retained austenite phase, and contains at least one of Nb, Ti, and V, the

finish rolling process. Application of the continuous rolling process is desirable also from the viewpoint of operational stability of hot rolling.

[0143] Then, a cold rolling step is conducted for the hot-rolled steel sheet. In the cold rolling step, the hot-rolled steel sheet is cold-rolled into a cold-rolled steel sheet. Any cold rolling conditions may be used so far as such conditions permit production of cold-rolled steel sheets with desired dimensions and shape, and no particular restriction is imposed. The reduction in cold rolling is preferably not less than 40%. With a reduction of less than 40%, uniform recrystallization barely occurs during the subsequent recrystallization-annealing step.

[0144] Then, the cold-rolled steel sheet is subjected to the recrystallization annealing step to convert the sheet into a cold-rolled annealed steel sheet. The recrystallization annealing is preferably carried out on a continuous annealing line. In the present invention, the recrystallization annealing is a heat treatment which includes heating and soaking the cold-rolled sheet in the dual phase region of ferrite and austenite in the temperature range between the A_{C1} transformation point and the A_{C3} transformation point, cooling the sheet, and retaining the sheet at a temperature in the range of 300 to 500°C for 30 to 1,200 seconds.

[0145] The heating and soaking temperature for recrystallization annealing is preferably within the dual phase region in the temperature range between the A_{C1} transformation point and the A_{C3} transformation point. The heating and soaking temperature of less than the A_{C1} transformation point leads to the formation a single ferrite phase. On the other hand, a high temperature exceeding A_{C3} transformation point results in coarsening of crystal grains, the formation of a single austenite phase, and a serious deterioration of press formability.

[0146] After the heating and soaking treatment, the sheet was cooled from the heating and soaking temperature and retained at a temperature in the range of 300 to 500°C for 30 to 1,200 seconds. The heating and soaking treatment and the subsequent retaining treatment facilitates the formation of a retained austenite phase of not less than 1%. When the temperature for the retaining treatment is less than 300°C, the composite structure of ferrite and martensite is formed. On the other hand, a temperature range exceeding 500°C leads to a ferrite/bainite composite structure or a ferrite/pearlite composite structure. In these cases, the retained austenite is barely formed.

[0147] In addition, a retention time of less than 30 seconds in the temperature range of 300 to 500°C cannot lead to the formation of the retained austenite structure. Also, the retention time exceeding 1,200 seconds cannot lead to the formation of the retained austenite structure, but leads to a ferrite/bainite composite structure. Therefore, the retention time in the temperature region of 300 to 500°C is preferably in the range of 30 to 1,200 seconds.

[0148] By the recrystallization annealing, a composite structure of a ferrite phase and a retained austenite phase is formed, whereby a high ΔTS can be obtained together with high ductility.

[0149] After the hot rolling, temper rolling with a reduction rate of not more than 10% may be applied for adjustments and other shape correction and, surface roughness control.

[0150] The cold-rolled steel sheet of the invention may be used as a steel sheet for processing and as a steel sheet for surface-treating. Surface treatments include galvanizing (including alloying), tin-plating and enameling. After galvanizing, the cold-rolled steel sheet of the present invention may be subjected to a special treatment to improve activity to chemical treatment, weldability, press formability, and corrosion resistance.

(3) Hot-dip galvanized steel sheet

[0151] The hot-dip galvanized steel sheet of the present invention will now be described.

[0152] The hot-dip galvanized steel sheet of the present invention has a composite structure comprising a primary phase consisting of a ferrite phase and a tempered martensite phase and a secondary phase containing retained austenite phase in a volume ratio of not less than 2%.

[0153] Note that the term "tempered martensite phase" in the present invention means a phase produced by heating a lath martensite. That is, the tempered martensite phase still maintains a fine internal structure of the lath martensite, after the heating (tempering). Furthermore, the tempered martensite phase is softened by heating (tempering), has enhanced deformability as compared with martensite, and is effective for improving ductility of the steel sheet. Note that the term "lath martensite" means martensite consisting of bundles of thin long platelike martensite crystals, which can be observed with an electron microscope.

[0154] In the hot-dip galvanized steel sheet of the present invention, the total volume ratio of the ferrite phase and the tempered martensite phase functioning as the primary phase is preferably not less than 50%. With a total volume ratio of the ferrite phase and the tempered phase of less than 50%, it is difficult to secure high ductility and press formability is decreased. When further enhanced ductility is required, the total volume ratio of the ferrite phase and the martensite phase functioning as the primary phase is preferably not less than 80%. For the purpose of making full use of advantages of the composite structure, the total of the ferrite phase and the tempered martensite phase is preferably not more than 98%. The ferrite phase constituting the primary phase preferably occupies not less than 30% by volume of the entire structure, and the tempered martensite phase preferably occupies not less than 20% by volume of the entire structure. With a volume ratio of the ferrite phase of less than 30%, or with a volume ratio of the tempered

present invention also include a steel making process using other deoxidizers, for example, Ti or Si, and steel sheets produced by such deoxidation methods are also included in the scope of the present invention. In this case, addition of Ca or REM to molten steel does not impair the features of the steel sheet of the present invention at all. Of course, steel sheets containing Ca or REM are included within the scope of the present invention.

N: not more than 0.02%

[0163] N is an element, which increases strength of a steel sheet through solid solution strengthening or strain age hardening, and is preferably contained in an amount of not less than 0.001%. An N content exceeding 0.02% causes an increase in the nitride content in the steel sheet, which causes serious deterioration of ductility and of press formability. The N content is, therefore, limited to not more than 0.02%. When further improvement of press formability is required, the N content is preferably not more than 0.01%.

Cu: 0.5 to 3.0%

[0164] Cu is an element, which remarkably increases strain age hardening of a steel sheet (increase in strength after pre-deformation/heat treatment), and is the most important element in the present invention. With a Cu content of less than 0.5%, an increase in tensile strength ΔTS of not less than 80 MPa cannot be obtained by changing the pre-deformation/heat treatment conditions. In the present invention, therefore, Cu should be contained in an amount of not less than 0.5%. With a Cu content exceeding 3.0%, however, the effect is saturated, leading to unfavorable economic effects. Furthermore, deterioration of press formability occurs, and the surface quality of the steel sheet is degraded. The Cu content is, therefore, limited within the range of 0.5 to 3.0%. In order to simultaneously achieve a higher ΔTS and excellent press formability, the Cu content is preferably within the range of 1.0 to 2.5%.

[0165] In the present invention, it is preferable that the composition containing Cu further contain, in weight percent, at least one of the following Groups A to C:

Group A: Ni: not more than 2.0%;
Group B: at least one of Cr and Mo: not more than 2.0% in total; and
Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

Group A: Ni: not more than 2.0%

[0166] Group A: Ni is an element effective for preventing surface defects produced by Cu contained in the steel sheet, and can be contained as required. The Ni content depends on the Cu content, and is preferably about a half the Cu content, more specifically, within the range of about 30 to about 80% of the Cu content. An Ni content exceeding 2.0% cannot give further enhancement in the effect because of saturation of the effect, leading to economic disadvantages, and causes deterioration of press formability. For these reasons, the Ni content is preferably limited to not more than 2.0%.

[0167] Group B: at least one of Cr and Mo: not more than 2.0% in total

[0168] Group B: Both Cr and Mo strengthen the steel sheet, like Mn, and can be contained as required. However, if at least one of Cr and Mo are contained in an amount exceeding 2.0% in total, press formability is impaired. The total content of Cr and Mo is preferably limited to not more than 2.0%. From the viewpoint of press formability, a Cr content is preferably not less than 0.1%, and an Mo content is preferably not less than 0.1%.

Group C: at least one of Nb, Ti, and V: not more than 0.2% in total

[0169] Group C: Nb, Ti, and V are carbide-forming elements and increase strength by fine dispersion of carbides, and can be selected and contained as required. However, if the total content of at least one of Nb, Ti, and V exceeds 0.2%, press formability is impaired. Thus, the total content of Nb, Ti and V is preferably limited to not more than 0.2%. The above-mentioned effect can be achieved at an Nb content of not less than 0.01%, at a Ti content of not less than 0.01%, and at a V content of not less than 0.01%.

[0170] In the present invention, in place of Cu, at least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0%, Cr, and W: 0.05 to 2.0% may be contained in an amount of not more than 2.0% in total.

At least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0%, in an amount of not more than 2.0% in total

[0171] In the present invention, all of Mo, Cr, and W, as well as Cu, are the most important elements, which remarkably

Slab reheating temperature: not less than 900°C

[0182] In the case of a steel slab containing Cu, the slab heating temperature is preferably the lowest possible to prevent surface defects caused by Cu. However, a heating temperature of less than 900°C causes an increase in the rolling load, thus increasing the risk of occurrence of a trouble during the hot rolling. Considering the increase in scale loss caused by accelerated oxidation, the slab heating temperature is preferably not more than 1,300°C. From the viewpoint of decreasing the slab heating temperature and preventing occurrence of troubles during hot rolling, use of a so-called sheet bar heater, which heats a sheet bar, is effective.

Finish rolling end temperature: not less than 700°C

[0183] At a finish rolling end temperature FDT of not less than 700°C, it is possible to obtain a uniform hot-rolled mother sheet structure which can give an excellent formability after cold rolling and recrystallization annealing. A finish rolling end temperature FDT of less than 700°C leads to a non-uniform structure of the hot-rolled mother sheet and a higher rolling load during hot rolling, thus increasing the risk of occurrence of troubles during hot rolling. Thus, the FDT for the hot rolling step is preferably not less than 700°C.

Coiling temperature: not more than 800°C

[0184] The coiling temperature CT is preferably not more than 800°C, and more preferably not less than 200°C. The CT exceeding 800°C tends to cause a decrease in yield as a result of an increased scale loss. With a CT of less than 200°C, the steel sheet shape is seriously impaired, and there is an increasing risk of occurrence of inconveniences in practical use.

[0185] The hot-rolled steel sheet suitably applicable in the present invention is preferably prepared by heating the slab to not less than 900°C, hot-rolling the heated slab at a finish rolling end temperature of not less than 700°C, and coiling the hot-rolled sheet at a coiling temperature of not less than 800°C, and preferably not less than 200°C.

[0186] In the above-mentioned hot rolling step, all or part of finish rolling may be lubrication rolling, which reduces the rolling load during the hot rolling. The lubrication rolling is effective also from the viewpoint of achieving a uniform steel sheet shape and a uniform material quality. The frictional coefficient on the lubrication rolling is preferably within the range of 0.25 to 0.10. It is desirable to connect neighboring sheet bars to each other to perform a continuous finish rolling process. Application of the continuous rolling process is desirable also from the viewpoint of operational stability of hot rolling.

[0187] The hot-rolled sheet with scales may be annealed to form an internal oxide layer at the surface of the steel sheet. The internal oxide layer, which prevents concentration of Si, Mn, and P at the surface, improves hot-dip galvanizing ability.

[0188] The hot-rolled sheet manufactured by the above-mentioned method may be used as an original sheet for plating. Alternatively, the hot-rolled sheet may be cold-rolled to form a cold-rolled sheet used as an original sheet for plating.

[0189] In the cold rolling step, any cold rolling condition may be used without particular restriction so far as such a condition permits production of cold-rolled steel sheets with desired dimensions and shapes. The reduction in cold rolling is preferably not less than 40%. A reduction of less than 40% inhibits uniform recrystallization during the subsequent primary heat treatment.

[0190] In the present invention, the above-mentioned steel sheet (hot-rolled sheet or cold-rolled sheet) is subjected to a primary heat treatment step including heating to a temperature of not less than the A_{C1} transformation point and rapid cooling.

[0191] Heating in the primary heat treatment, the steel sheet is preferably held at a temperature of not less than A_{C1} transformation point, more preferably not less than (A_{C3} transformation point - 50°C), and most preferably not less than A_{C3} transformation point. After heating, the steel sheet is preferably rapidly cooled to a temperature of not more than the Ms point at a cooling rate of not less than 10°C/second. During the primary heat treatment step, lath martensite is produced in the steel sheet. In the present invention, the most important point is to form lath martensite during the primary heat treatment step. Unless the lath martensite is formed in the steel sheet, it is difficult to form a secondary phase containing retained austenite in the subsequent steps.

[0192] When a hot-rolled steel sheet, subjected to final hot rolling at a temperature of not less than (Ar_3 transformation point - 50°C), is used as an original sheet for plating, the primary heat treatment step can be substituted the steel sheet for rapidly cooling to a temperature of not less than Ms point at a cooling rate of not less than 10°C/second during cooling after the final hot rolling.

[0193] Then, the steel sheet containing lath martensite formed during the above-described primary heat treatment is subjected to a secondary heat treatment step for heating to and holding at a temperature in the range of A_{C1} trans-

Table 1

STEEL NO.	COMPOSITION (wt.%)										
	C	Si	Mn	P	S	Al	N	Cu	Ni	Cr, Mo,	Nb, Ti, V
A	0.09	1.45	1.05	0.01	0.003	0.034	0.002	1.52	-	-	-
B	0.12	1.50	1.20	0.01	0.002	0.030	0.002	1.43	0.65	Mo: 0.32	-
C	0.10	1.48	1.35	0.01	0.002	0.028	0.002	1.25	0.52	Cr: 0.53	-
D	0.15	1.53	1.45	0.01	0.003	0.033	0.002	1.33	0.44	-	Nb: 0.01, Ti: 0.01, V: 0.01
E	0.12	1.48	1.55	0.01	0.005	0.032	0.002	0.15	-	-	-
F	0.11	1.50	1.08	0.01	0.004	0.032	0.002	0.68	-	-	-
G	0.13	1.52	1.22	0.01	0.004	0.030	0.002	0.98	-	-	-
H	0.12	1.42	1.22	0.01	0.003	0.033	0.002	1.55	0.62	-	-
I	0.11	1.52	1.52	0.01	0.003	0.031	0.002	1.49	-	Cr: 0.15, Mo: 0.12	-
J	0.13	1.43	1.48	0.01	0.003	0.028	0.002	1.43	-	Mo: 0.21	-
K	0.15	1.58	1.05	0.01	0.003	0.030	0.002	1.52	-	-	Nb: 0.01
L	0.14	1.60	1.21	0.01	0.003	0.028	0.002	1.48	-	Cr: 0.11	Ti: 0.01

[0205] For the resulting hot-rolled steel strip (hot-rolled steel sheet), the microstructure, tensile properties, strain age hardenability, and hole expanding property were determined. Press formability was evaluated in terms of elongation El (ductility), TS × El balance and hole expanding ratio λ. Test methods were as follows.

(1) Microstructure

[0206] A test piece was sampled from each of the resultant hot-rolled sheets, and the microstructure of the cross-section (section C) perpendicular to the rolling direction of the steel sheet was observed with an optical microscope and a scanning electron microscope. The volume ratios of the ferrite phase, the bainite phase, and the martensite phase in the steel sheet were determined with an image analyzer using a photograph of the cross-sectional structure at a magnification of 1,000. The volume ratios of the retained austenite phase were determined by polishing the steel sheet to the central plane in the thickness direction, and by measuring diffraction X-ray intensities at the central plane. Mo Kα-rays were used as incident X-rays, the ratios of the diffraction X-ray strengths of the planes {200}, {220} and {311} of the retained austenite phase to the diffraction X-ray strengths of the planes {110}, {200} and {211} of the ferrite phase, respectively, were determined, and the volume ratio of the retained austenite was determined from the average of these ratios.

(2) Tensile properties

[0207] JIS No. 5 tensile test pieces were sampled from the resultant hot-rolled sheets, and a tensile test was carried out in accordance with JIS Z 2241 to determine the yield strength YS, the tensile strength TS, and the elongation El.

(3) Strain age hardenability

[0208] JIS No. 5 test pieces were sampled in the rolling direction from the resultant hot-rolled steel sheets. A plastic deformation of 5% was applied as a pre-deformation (tensile prestrain). After a heat treatment at 250°C for 20 minutes, a tensile test was carried out to determine tensile properties (yield stress YS_{TH} and tensile strength TS_{HT}) and to calculate $\Delta YS = YS_{TH} - YS$, and $\Delta TS = TS_{HT} - TS$, wherein YS_{TH} and TS_{HT} were yield stress and tensile strength after the pre-deformation/heat treatment, and YS and TS were yield stress and tensile strength of the hot-rolled steel sheets.

(4) Hole expanding property

[0209] A hole was formed by punching a test piece sampled from the resultant hot-rolled sheet in accordance with Japan Iron and Steel Federation Standard (JFS T 1001-1996) with a punch having a diameter of 10 mm. Then, the hole was expanded with a conical punch having a vertical angle of 60° so that burrs were produced on the outside until cracks passing through the thickness form, thereby determining the hole expanding ratio λ. The hole expanding ratio λ was calculated by the formula: λ (%) = $\{(d - d_0)/d_0\} \times 100$, where d_0 is initial hole diameter (punch diameter), and d is inner hole diameter upon occurrence of cracks.

[0210] The results are shown in Table 3.

[0211] All Examples according to the present invention show a high elongation El, a high strength/ductility balance (TS × El), and a high hole expanding ratio λ, suggesting excellent stretch flanging formability. In addition, all Examples according to the present invention show a very large ΔTS, suggesting that these samples had excellent strain age hardenability. Comparative Examples outside the scope of the present invention, in contrast, suggest that the samples have a low elongation El, a small hole expanding ratio λ, a low ΔTS, and decreased press formability and strain age hardenability.

(Example 2)

[0212] Molten steels having the compositions shown in Table 4 were made in a converter and cast into steel slabs by a continuous casting process. Each of these steel slabs were reheated, and hot-rolled under conditions shown in Table 5 into a hot-rolled steel strip (hot-rolled sheet) having a thickness of 2.0 mm. The hot-rolled steel strip was temper-rolled at a reduction of 1.0%.

Table 5

STEEL SHEET NO.	STEEL NO.	SLAB REHEATING TEMP. SRT °C	HOT ROLLING - COOLING AFTER ROLLING										
			FINISH ROLLING END TEMP. FDT °C	TIME BEFORE START COOLING S	FORCED COOLING		ISOTHERMAL HOLDING		SLOW COOLING TREATMENT			COOLING RATE BEFORE COILING °C/s	COILING TEMP. CT °C
					COOLING RATE °C/s	STOP TEMP. °C	TEMP. °C	HOLDING TIME S	INITIAL TEMP. °C	COOLING RATE °C/s	STOP TEMP. °C		
2-1	2A	1250	850	0.5	90	680	680	5	-	-	-	80	450
2-2	2B	1250	850	0.5	80	710	710	5	-	-	-	60	450
2-3		1250	850	0.3	30	-	-	-	700	10	670	30	600
2-4		1250	850	0.5	30	-	-	-	690	10	660	20	450
2-5	2C	1250	850	0.1	70	680	680	5	-	-	-	60	450
2-6	2D	1250	850	0.5	80	680	680	5	-	-	-	80	450
2-7	2E	1250	850	0.5	80	700	700	5	-	-	-	80	450
2-8	2F	1250	850	0.5	70	690	690	5	-	-	-	70	450
2-9	2G	1250	850	0.5	80	680	680	5	-	-	-	80	450
2-10	2H	1250	850	0.3	60	680	680	5	-	-	-	60	450
2-11	2I	1250	850	0.5	80	700	700	5	-	-	-	60	450
2-12	2J	1250	850	0.5	80	700	700	5	-	-	-	60	450
2-13	2K	1250	850	0.1	70	690	690	5	-	-	-	70	450
2-14	2L	1250	850	0.5	60	680	680	5	-	-	-	60	450

Table 6

STEEL SHEET NO.	STEEL NO.	MICROSTRUCTURE				HOT-ROLLED SHEET PROPERTIES				PROPERTIES AFTER PRE-DEFORMATION - HEAT TREATMENT		STRAIN AGE HARDENING PROPERTIES		HOLE EXPANSION	REMARKS
		PRIMARY PHASE	SECONDARY PHASE			TENSILE PROPERTIES								HOLE EXPANDING RATIO λ %	
		F VOLUME RATIO %	A VOLUME RATIO %	OTHER PHASES KIND*	VOLUME RATIO %	YS (MPa)	TS (MPa)	El%	TS×El MPa%	YS_{HT} MPa	TS_{HT} MPa	ΔYS MPa	ΔTS MPa		
2-1	2A	76	8	B, M	24	460	610	35	21350	695	760	235	150	135	EXAMPLE
2-2	2B	79	9	B, M	21	480	640	33	21120	730	800	250	160	140	EXAMPLE
2-3		76	-	P	24	650	710	15	10650	700	730	50	20	70	COMP.EX.
2-4		75	-	P, B	25	590	650	14	9100	635	665	45	15	65	COMP.EX.
2-5	2C	76	9	B, M	24	480	630	34	21420	715	785	235	155	140	EXAMPLE
2-6	2D	78	8	B, M	22	490	650	33	21450	725	810	235	160	135	EXAMPLE
2-7	2E	80	7	B, M	20	390	510	42	21420	620	670	230	160	130	EXAMPLE
2-8	2F	81	9	B, M	19	450	590	36	21240	660	730	210	140	135	EXAMPLE
2-9	2G	79	10	B, M	21	450	600	36	21600	570	630	120	30	65	COMP.EX.
2-10	2H	78	10	B, M	22	480	630	34	21420	715	785	235	155	130	EXAMPLE
2-11	2I	80	8	B, M	20	460	610	35	21350	695	760	235	150	135	EXAMPLE
2-12	2J	79	9	B, M	21	450	590	36	21240	660	730	210	140	130	EXAMPLE
2-13	2K	80	9	B, M	20	460	600	35	21000	670	750	200	150	140	EXAMPLE
2-14	2L	81	8	B, M	19	470	620	34	21080	670	780	200	160	135	EXAMPLE

* F: FERRITE, A: AUSTENITE, M: MARTENSITE, P: PEARLITE, B: BAINITE

Table 7

STEEL NO.	COMPOSITION (wt.%)														TRANSFORMATION POINT (°C)	
	C	Si	Mn	P	S	Al	N	Cu	Ni	Cr	Mo	Nb	Ti	V	Ac1	Ac3
3A	0.10	1.20	1.42	0.01	0.003	0.032	0.002	1.51	-	-	-	-	-	-	725	875
3B	0.11	1.10	1.51	0.01	0.002	0.033	0.002	1.45	0.63	-	0.11	-	-	-	715	875
3C	0.11	1.32	1.33	0.01	0.004	0.025	0.002	1.20	0.52	0.12	-	-	-	-	725	880
3D	0.10	1.06	1.48	0.01	0.003	0.022	0.002	1.39	0.43	-	-	0.01	0.01	0.01	720	870
3E	0.09	1.25	1.36	0.01	0.004	0.029	0.002	0.22	-	-	-	-	-	-	730	860
3F	0.10	1.08	1.45	0.01	0.001	0.030	0.002	0.75	-	-	-	-	-	-	720	880
3G	0.11	1.15	1.52	0.01	0.002	0.033	0.002	0.96	-	-	-	-	-	-	725	875
3H	0.10	1.10	1.55	0.01	0.002	0.025	0.002	1.22	0.66	-	-	-	-	-	730	875
3I	0.11	1.09	1.48	0.01	0.001	0.033	0.002	1.36	-	-	0.10	-	-	-	725	860
3J	0.11	1.12	1.62	0.01	0.002	0.029	0.001	1.42	-	0.10	-	-	-	-	730	880
3K	0.10	1.25	1.39	0.01	0.002	0.032	0.002	1.38	-	-	-	0.01	-	-	720	870
3L	0.09	1.10	1.45	0.01	0.003	0.025	0.002	1.29	-	-	-	-	0.01	-	725	865
3M	0.10	1.35	1.50	0.01	0.002	0.030	0.002	1.44	-	-	-	-	-	0.01	730	875
3N	0.11	1.26	1.46	0.01	0.001	0.028	0.001	1.33	0.52	0.12	0.11	0.01	0.01	0.01	725	865

the thickness direction and by measuring diffraction X-ray intensities at the central plane. The incident X-ray, the planes of the ferrite phase, and the planes of retained austenite used were the same as those in Example 1.

(2) Tensile properties

[0219] JIS No. 5 tensile test pieces were sampled from the resultant steel strips in the direction perpendicular to the rolling direction, and a tensile test was carried out, as in Example 1, in accordance with JIS Z 2241 to determine yield strength YS, tensile strength TS, and elongation El.

(3) Strain age hardenability

[0220] JIS No. 5 test pieces were sampled in the direction perpendicular to the rolling direction from the resultant steel strips (cold-rolled annealed sheets). A plastic deformation of 5% was applied as a pre-deformation (tensile pre-strain), as in Example 1. After a heat treatment at 250°C for 20 minutes, a tensile test was carried out to determine tensile properties (yield stress YS_{HT}, and tensile strength TS_{HT}) and to calculate $\Delta YS = YS_{HT} - YS$, and $\Delta TS = TS_{HT} - TS$, wherein YS_{HT} and TS_{HT} were yield stress and tensile strength after the pre-deformation -heat treatment, and YS and TS were yield stress and tensile strength of the steel strips (cold-rolled annealed sheets).

(4) Hole expanding property

[0221] A hole was formed by punching a test piece sampled from the resultant steel strip in accordance with Japan Iron and Steel Federation Standard JFS T 1001-1996 with a punch having a diameter of 10 mm. Then, the hole was expanded with a conical punch having a vertical angle of 60° so that burrs were produced on the outside until cracks passing through the thickness form, thereby determining the hole expanding ratio λ, as in Example 1. The results are shown in Table 9.

[0222] All Examples according to the present invention are cold-rolled steel sheets having a high elongation El, a high strength-elongation balance TS × El, a high hole expanding ratio λ, and excellent press formability including stretch flanging formability. In addition, Examples according to the present invention each show a very large ΔTS, suggesting that the samples have excellent strain age hardenability. Comparative Examples outside the scope of the present invention, in contrast, suggest that the samples each have a low elongation El, a low TS × El, a small hole expanding ratio λ, a low ΔTS, and decreased press formability and strain age hardenability.

(Example 4)

[0223] Molten steels having the compositions shown in Table 10 were made in a converter and cast into steel slabs by a continuous casting process. Each of these steel slabs were reheated to 1,250°C, and hot-rolled by a hot rolling step of hot rolling with a finish rolling end temperature of 900°C and a coiling temperature of 600°C into a hot-rolled steel strip (hot-rolled sheet) having a thickness of 4.0 mm. Then, the hot-rolled steel strip (hot-rolled sheet) was subjected to a cold rolling step of pickling and cold-rolling into a cold rolled steel strip (cold-rolled sheet) having a thickness of 1.2 mm. Thereafter, the cold-rolled steel strip (cold-rolled sheet) was subjected to recrystallization annealing step comprising a heating and soaking treatment and a subsequent retaining treatment under the conditions shown in Table 11 on a continuous annealing line to obtain cold-rolled annealed sheet. The resultant steel strip (cold-rolled annealed sheet) was further temper-rolled at an reduction of 0.8%.

Table 11

STEEL SHEET NO.	STEEL NO.	SLAB REHEATING TEMP. (°C)	HOT ROLLING STEP		COLD ROLLING STEP	RECRYSTALLIZATION ANNEALING STEP		
			FINISH ROLLING END TEMP. FDT °C	COILING TEMP. CT °C	COLD ROLLING REDUCTION %	HEATING SOAKING TREATMENT	RETAINING TREATMENT	
						HEATING SOAKING TEMP. (°C)	TEMP. (°C)	RETENTION TIME (s)
4-1	4A	1250	900	600	70	800	400	300
4-2	4B	1250	900	600	70	800	400	300
4-3		1250	900	600	70	<u>980</u>	-	-
4-4		1250	900	600	70	<u>680</u>	400	300
4-5	4C	1250	900	600	70	800	400	300
4-6	4D	1250	900	600	70	800	400	300
4-7	4E	1250	900	600	70	800	400	300
4-8	4F	1250	900	600	70	800	400	300
4-9	4G	1250	900	600	70	800	400	300
4-10	4H	1250	900	600	70	800	400	300
4-11	4I	1250	900	600	70	800	400	300
4-12	4J	1250	900	600	70	800	400	300
4-13	4K	1250	900	600	70	800	400	300
4-14	4L	1250	900	600	70	800	400	300
4-15	4A	1250	900	600	70	800	<u>250</u>	300
4-16		1250	900	600	70	800	<u>550</u>	300

[0224] A test piece was sampled from the resultant steel strip, and the microstructure, the tensile properties, the strain age hardenability, and the hole expanding property were investigated, as in Example 3.

[0225] The results are shown in Table 12.

[0226] All Examples according to the present invention show a high elongation El, a high strength-ductility balance TS × El, and a high hole expanding ratio λ, suggesting that the samples have excellent press formability including stretch flanging formability. In addition, Examples according to the present invention show a very large ΔTS, suggesting that the samples have excellent strain age hardenability. Comparative Examples outside the scope of the present invention, in contrast, suggest that the samples have a low elongation El, a low TS x El, a small hole expanding ratio λ, a low ΔTS, and decreased press formability and strain age hardenability.

(Example 5)

[0227] Molten steels having the compositions shown in Table 13 were made in a converter and cast into steel slabs by a continuous casting process. These slabs were hot-rolled under the conditions shown in Table 14 into hot-rolled steel strips (hot-rolled sheets).

[0228] After pickling, each of these hot-rolled steel strips (hot-rolled sheets) was subjected to a primary heat treatment step on a continuous annealing line (CAL) under the conditions shown in Table 14 and a secondary heat treatment step on a continuous hot-dip galvanizing line (CGL) under the conditions shown in Table 14. Then, the sheet was subjected to a hot-dip galvanizing treatment step of performing a hot-dip galvanizing which forms a hot-dip galvanizing layer on the surfaces of the steel sheet. Then, an alloying treatment step of alloying the hot-dip galvanizing layer was applied under the conditions shown in Table 14. Some of the steel sheets were left as hot-dip galvanized.

[0229] After further pickling, the hot-rolled steel strip (hot-rolled sheet) obtained by the above-mentioned hot rolling was subjected to a cold rolling step under the conditions shown in Table 14 into a cold-rolled steel strip (cold-rolled sheet). Then, the cold-rolled steel strip (cold-rolled sheet) was subjected to a primary heat treatment step on a continuous annealing line (CAL) under the conditions shown in Table 14. After a secondary heat treatment step on the continuous hot-dip galvanizing line (CGL) under the conditions shown in Table 14, a hot-dip galvanizing treatment step was performed. Then, an alloying treatment step was performed under the conditions shown in Table 14. Some of the steel sheets were left as hot-dip galvanized.

[0230] Prior to the secondary heat treatment step on the continuous hot-dip galvanizing line (CGL), some of the steel sheets after the primary heat treatment step were subjected to a pickling treatment shown in Table 14. The pickling treatment was carried out in a pickling bath on the entry side of the CGL.

[0231] The galvanizing bath temperature was within the range of 460 to 480°C, and the temperature of the steel sheet to be dipped was within the range of the galvanizing bath temperature to (bath temperature + 10°C). In the alloying treatment, the sheet was reheated within the temperature range of 480 to 540°C, and held at the temperature for 15 to 28 seconds. The cooling rate after the alloying treatment was 10°C/second. The plated steel sheet was further temper rolled at a reduction of 1.0%.

Table 14

STEEL SHEET NO.	STEEL NO.	SLAB REHEATING TEMP. (°C)	HOT ROLLING STEP			COLD ROLLING STEP		PRIMARY HEAT TREATMENT STEP			PICKLING TREATMENT	SECONDARY HEAT TREATMENT STEP			HOT-DIP GALVANIZING		ALLOYING TREATMENT STEP		TEMPER ROLLING
			FINISH ROLLING END TEMP. FDT °C	COILING TEMP. CT °C	FINAL THICK-NESS mm	COLD ROLLING REDUCTION %	FINAL THICK-NESS mm	LINE	HEATING TEMP. °C	COOLING RATE °C/s		KIND OF LINE	HEATING TEMP. °C	COOLING RATE* °C/s	KIND OF LINE	COOLING RATE AFTER GALVANIZING **°C/s		TEMP. °C	REDUCTION %
5-1	5A	1250	850	600	1.2	-	-	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-2	5B	1250	850	600	1.2	-	-	CAL	880	20	-	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-3											YES	CGL	780	20	CGL	10	ALLOYING	500	1.0
5-4												CGL	980	20	CGL	10	ALLOYING	500	1.0
5-5												CGL	650	20	CGL	10	ALLOYING	500	1.0
5-6	5C	1250	850	600	1.2	-	-	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-7	5D	1250	850	600	1.2	-	-	CAL	880	20	YES	CGL	820	20	CGL	10	ALLOYING	500	1.0
5-8	5E	1250	850	600	1.2	-	-	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-9	5F	1250	850	600	1.2	-	-	CAL	880	20	YES	CGL	780	20	CGL	10	NON-ALLOYING	-	1.0
5-10	5G	1250	850	600	1.2	-	-	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-11	5A	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-12	5B	1250	850	600	4.0	70	1.2	CAL	880	20	-	CGL	820	20	CGL	10	ALLOYING	500	1.0
5-13								CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-14								CAL	880	20	YES	CGL	980	20	CGL	10	ALLOYING	500	1.0
5-15								CAL	880	20	YES	CGL	680	20	CGL	10	ALLOYING	500	1.0
5-16	5C	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-17	5D	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	NON-ALLOYING	-	1.0
5-18	5E	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	780	20	CGL	10	ALLOYING	500	1.0
5-19	5F	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-20	5G	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	820	20	CGL	10	ALLOYING	500	1.0
5-21	5H	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-22	5I	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-23	5J	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-24	5K	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-25	5L	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-26	5M	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0
5-27	5N	1250	850	600	4.0	70	1.2	CAL	880	20	YES	CGL	800	20	CGL	10	ALLOYING	500	1.0

*) COOLING RATE UNTIL 480°C **) COOLING RATE UNTIL 300°C

Table 15

STEEL SHEET NO.	STEEL NO.	MICROSTRUCTURE						PLATED SHEET PROPERTIES				PROPERTIES AFTER PRE-DEFORMATION - HEAT TREATMENT		STRAIN AGE HARDENING PROPERTIES		HOLE EXPANSION	REMARKS
		PRIMARY PHASE			SECONDARY PHASE			TENSILE PROPERTIES								HOLE EXPANDING RATIO λ %	
		FERRITE VOLUME RATIO %	TEMPERED MARTENSITE VOLUME RATIO %	VOLUME RATIO %	KIND*	RETAINED AUSTENITE VOLUME RATIO %	VOLUME RATIO %	YS (MPa)	TS (MPa)	El (%)	TS×El (MPa%)	YS_{HT} (MPa)	TS_{HT} (MPa)	ΔYS (MPa)	ΔTS (MPa)		
5-1	5A	57	35	92	A, B,	5	8	470	620	34	21080	700	775	230	155	140	EXAMPLE
5-2	5B	52	40	92	A, B,	4	8	480	640	33	21120	725	805	245	165	135	EXAMPLE
5-3		51	40	91	A, B,	5	9	470	620	34	21080	710	785	240	165	135	EXAMPLE
5-4		0	0	0	M, P, B	0	100	670	710	11	7810	710	740	40	30	65	COMP.EX.
5-5		60	40	100	-	0	0	620	650	12	7800	650	675	30	25	130	COMP.EX.
5-6	5C	58	35	93	A,B	4	7	470	630	34	21420	710	785	240	155	135	EXAMPLE
5-7	5D	57	35	92	A,B	5	8	490	650	33	21450	725	805	235	155	130	EXAMPLE
5-8	5E	53	40	93	A,B	7	7	380	510	42	21420	480	530	100	20	60	COMP.EX.
5-9	5F	37	55	92	A,B	4	8	430	570	37	21030	650	720	220	150	140	EXAMPLE
5-10	5G	53	40	93	A,B	5	7	450	590	36	21240	675	745	225	155	135	EXAMPLE
5-11	5A	57	35	92	A,B	7	8	470	630	34	21420	715	790	245	160	145	EXAMPLE
5-12	5B	52	40	92	A,B	5	8	500	660	32	21120	750	830	250	170	140	EXAMPLE
5-13		53	40	93	A,B	6	7	480	640	33	21120	730	810	250	170	140	EXAMPLE
5-14		0	0	0	M,P,B	0	100	680	720	12	8640	720	750	40	30	70	COMP.EX.
5-15		65	35	100	-	0	0	620	660	11	7260	650	685	30	25	60	COMP.EX.
5-16	5C	52	40	92	A,B	4	8	490	650	33	21450	730	810	240	160	140	EXAMPLE
5-17	5D	53	40	93	A,B	5	7	500	660	32	21120	735	815	235	155	135	EXAMPLE
5-18	5E	48	45	93	A,B	4	7	390	520	41	21320	490	540	100	20	60	COMP.EX.
5-19	5F	44	50	94	A,B	5	6	440	580	37	21460	655	725	215	145	135	EXAMPLE
5-20	5G	57	35	92	A,B	5	8	450	600	35	21000	675	750	225	150	140	EXAMPLE
5-21	5H	51	40	91	A,B	5	9	445	590	35	20650	680	755	235	165	130	EXAMPLE
5-22	5I	55	35	90	A,B	5	10	460	610	34	20740	695	770	235	160	135	EXAMPLE
5-23	5J	52	40	92	A,B	4	8	450	600	35	21000	680	755	230	155	130	EXAMPLE
5-24	5K	53	40	93	A,B	5	7	470	620	34	21080	710	780	240	160	130	EXAMPLE
5-25	5L	56	35	91	A,B	6	9	475	630	33	20790	720	795	245	165	135	EXAMPLE
5-26	5M	60	30	90	A,B	5	10	460	610	34	20740	695	770	235	160	130	EXAMPLE
5-27	5N	52	40	92	A,B	4	8	455	600	35	21000	680	755	225	155	130	EXAMPLE

*) M: MARTENSITE, P: PEARLITE, B: BAINITE, A: RETAINED AUSTENITE

Table 16

STEEL NO.	COMPOSITION (wt.%)									TRANSFORMATION POINT (°C)	
	C	Si	Mn	P	S	Al	N	Cr, Mo, W	Nb, Ti, V	Ac1	Ac3
6A	0.07	0.77	2.00	0.01	0.003	0.033	0.002	Cr:0.20,Mo:0.43	-	715	870
6B	0.08	0.55	2.22	0.01	0.001	0.033	0.002	Mo:0.33	Nb:0.04,V:0.05	720	865
6C	0.08	0.75	1.80	0.01	0.004	0.020	0.002	Mo:0.48	Nb:0.05,Ti:0.03	725	880
6D	0.09	0.63	1.98	0.01	0.005	0.025	0.002	W:0.54	-	715	865
6E	0.07	0.65	2.02	0.01	0.003	0.033	0.002	Mo:0.36	Ti:0.05	715	875
6F	0.08	0.70	1.90	0.01	0.005	0.035	0.002	Cr:0.50	Nb:0.05	715	865
6G	0.07	0.58	2.08	0.01	0.004	0.032	0.002	-	-	715	865
6H	0.08	0.75	2.22	0.01	0.004	0.022	0.002	Mo:0.35	-	715	870
6I	0.08	0.77	1.98	0.01	0.003	0.032	0.002	Cr:0.25	-	710	860
6J	0.07	0.68	2.05	0.01	0.002	0.035	0.002	Mo:0.15, Cr:0.10, W:0.11	-	720	865
6K	0.09	0.70	1.98	0.01	0.001	0.028	0.002	Mo:0.25, Cr:0.10	V:0.05	715	865

[0240] A piece was sampled from the resultant hot-dip galvanized steel strip, and the microstructure, the tensile properties, the strain age hardenability, and the bore expanding property were investigated, as in Example 5.
[0241] The results are shown in Table 18.

[0242] All Examples according to the present invention show a high elongation El and a high bore expanding ratio λ, suggesting that the examples are hot-dip galvanized steel sheets having excellent press formability. In addition, all Examples according to the present invention show a very large ΔTS, suggesting that the samples are steel sheets having excellent strain age hardenability. Comparative Examples outside the scope of the invention, in contrast, suggest that the samples are steel sheets having a low elongation El, a low λ, a low ΔTS, and decreased press formability and strain age hardenability.

[0243] According to the present invention, it is possible to stably manufacture steel sheets (hot-rolled steel sheets, cold-rolled steel sheets and hot-dip galvanized steel sheets) in which the tensile strength is remarkably increased through a heat treatment applied after press forming while maintaining excellent press formability, giving industrially remarkable effects. When applying a steel sheet of the present invention to automotive parts, there are available advantages of easy press forming, high and stable parts properties after completion, and sufficient contribution to the weight reduction of the automobile body.

Claims

1. A high-ductility steel sheet excellent in press formability and in strain age hardenability as represented by a ΔTS of not less than 80 Mpa, comprising a composite structure containing a primary phase containing a ferrite phase and a secondary phase containing a retained austenite phase in a volume ratio of not less than 1%.

2. A high-ductility steel sheet according to Claim 1, wherein the steel sheet is a hot-rolled steel sheet, and the primary phase containing the ferrite phase is a ferrite phase.

3. A high-ductility steel sheet according to Claim 2, wherein the hot-rolled steel sheet has a composition comprising, in weight percent, C: 0.05 to 0.20%, Si: 1.0 to 3.0%, Mn: not more than 3.0%, P: not more than 0.10%, S: not more than 0.02%, Al: not more than 0.30%, N: not more than 0.02%, and Cu: 0.5 to 3.0%, and the balance Fe and incidental impurities.

4. A high-ductility steel sheet according to Claim 3, the composition further comprising, in weight percent, at least one of the following Groups A to C:

   Group A: Ni: not more than 2.0%;
   Group B: at least one of Cr and Mo: not more than 2.0% in total; and
   Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

5. A high-ductility steel sheet according to Claim 2, wherein the hot-rolled steel sheet has a composition comprising, in weight percent, C: 0.05 to 0.20%, Si: 1.0 to 3.0%, Mn: not more than 3.0%, P: not more than 0.10%, S: not more than 0.02%, Al: not more than 0.30%, N: not more than 0.02%, at least one of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0%, not more than 2.0% in total, and the balance Fe and incidental impurities.

6. A high-ductility steel sheet according to Claim 5, the composition further comprising, in weight percent, at least one of Nb, Ti, and V, in an amount of not more than 2.0% in total.

7. A method for manufacturing a high-ductility hot-rolled steel sheet excellent in press formability and in strain age hardenability as represented by a ΔTS of not less than 80 MPa, comprising the steps of:

   hot-rolling a steel slab having a composition comprising, in weight percent, C: not more than 0.20%, Si: 1.0 to 3.0%, Mn: not more than 3.0%, P: not more than 0.10%, S: not more than 0.02%, Al: not more than 0.30%, N: not more than 0.02%, and Cu: 0.5 to 3.0%, into a hot-rolled steel sheet having a prescribed thickness, the hot rolling step including finish-rolling at a finish rolling end temperature of 780 to 980°C;
   cooling the finish-rolled steel sheet to a temperature in the range of 620 to 780°C within 2 seconds at a cooling rate of not less than 50°C/second;
   holding the sheet at the temperature in the range of 620 to 780°C for 1 to 10 seconds, or slowly cooling the sheet at a cooling rate of not more than 20°C/second;
   cooling the sheet at a cooling rate of not less than 50°C/second to a temperature of 300 to 500°C; and coiling the sheet.

8. A method for manufacturing a high-ductility hot-rolled steel sheet excellent in press formability and in strain age

comprising, in weight percent, at least one selected from the following Groups A to C:

Group A: Ni: not more than 2.0%;
Group B: at least one of Cr and Mo: not more than 2.0% in total; and
Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

19. A method for manufacturing a high-ductility cold-rolled steel sheet according to Claim 17, wherein the steel slab is replaced with a steel slab having a composition containing, in weight percent, C: not more than 0.20%, Si: not more than 2.0%, Mn: not more than 3.0%, P: not more than 0.10%, S: not more than 0.02%, Al: not more than 0.3%, N: not more than 0.02%, and at least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0% in a total amount of not more than 2.0%.

20. A method for manufacturing a high-ductility cold-rolled steel sheet according to Claim 19, the composition further comprising, in weight percent, at least one of Nb, Ti, and V in a total amount of not more than 2.0%.

21. A method for manufacturing a high-ductility cold-rolled steel sheet according to any one of Claims 17 to 20, wherein the hot-rolling step includes heating the steel slab at a temperature of not less than 900°C, rolling the slab at a finish rolling end temperature of not less than 700°C, and coiling the hot-rolled steel sheet at a coiling temperature of not more than 800°C.

22. A method for manufacturing a cold-rolled steel sheet according to any one of Claims 17 to 21, wherein all or part of the hot rolling is lubrication rolling.

23. A high-ductility hot-dip galvanized steel sheet comprising a hot-dip galvanizing layer or an alloyed hot-dip galvanizing layer formed on the surface of the high-ductility steel sheet according to any one of Claims 1 to 6.

24. A high-ductility hot-dip galvanized steel sheet comprising a hot-dip galvanizing layer or an alloyed hot-dip galvanizing layer formed on the surface of the high-ductility steel sheet according to any one of Claims 12 to 16.

25. A high-ductility steel sheet according to Claim 1, wherein the steel sheet is a hot-dip galvanized steel sheet having a hot-dip galvanizing layer or an alloyed hot-dip galvanizing layer formed on a surface of the steel sheet, and the primary phase containing a ferrite phase comprises a ferrite phase and a tempered martensite phase.

26. A high-ductility steel sheet according to Claim 25, wherein the steel sheet has a composition comprising, in weight percent, C: not more than 0.20%, Si: not more than 2.0%, Mn: not more than 3.0%, P: not more than 0.1%, S: not more than 0.02%, Al: not more than 0.3%, N: not more than 0.02%, Cu: 0.5 to 3.0%, and the balance Fe and incidental impurities.

27. A high-ductility steel sheet according to Claim 26, the composition further comprising, in weight percent, at least one of the following Groups A to C:

Group A: Ni: not more than 2.0%;
Group B: at least one of Cr and Mo: not more than 2.0% in total; and
Group C: at least one of Nb, Ti, and V: not more than 0.2% in total.

28. A high-ductility steel sheet according to Claim 25, wherein the steel sheet has a composition comprising, in weight percent, C: not more than 0.20%, Si: not more than 2.0%, Mn: not more than 3.0%, P: not more than 0.1%, S: not more than 0.02%, Al: not more than 0.3%, N: not more than 0.02%, at least one selected from the group consisting of Mo: 0.05 to 2.0%, Cr: 0.05 to 2.0% and W: 0.05 to 2.0% in a total amount of not more than 2.0%, and the balance Fe and incidental impurities.

29. A high-ductility steel sheet according to Claim 28, the composition further comprising, in weight percent, at least one of Nb, Ti, and V in a total amount of not more than 2.0%.

30. A method of manufacturing of a high-ductility hot-dip galvanized steel sheet excellent in press formability and in strain age hardenability as typically represented by a ΔTS of not less than 80 MPa, comprising:

a primary heat-treating step of heating a steel sheet to a temperature of not less than the A_{C1} transformation

Fig. 1

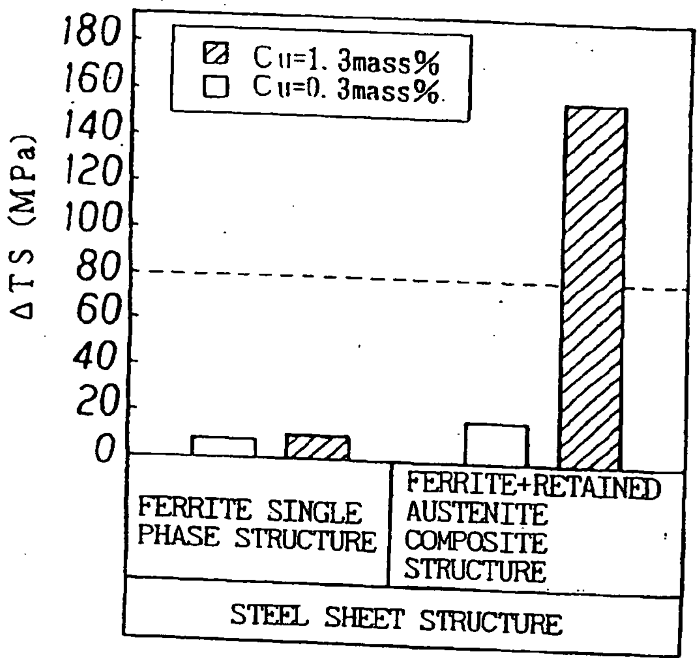

Cu=1. 3mass%
Cu=0. 3mass%
ΔTS (MPa)
180
160
140
120
100
80
60
40
20
0
FERRITE SINGLE PHASE STRUCTURE
FERRITE+RETAINED AUSTENITE COMPOSITE STRUCTURE
STEEL SHEET STRUCTURE

Fig. 2

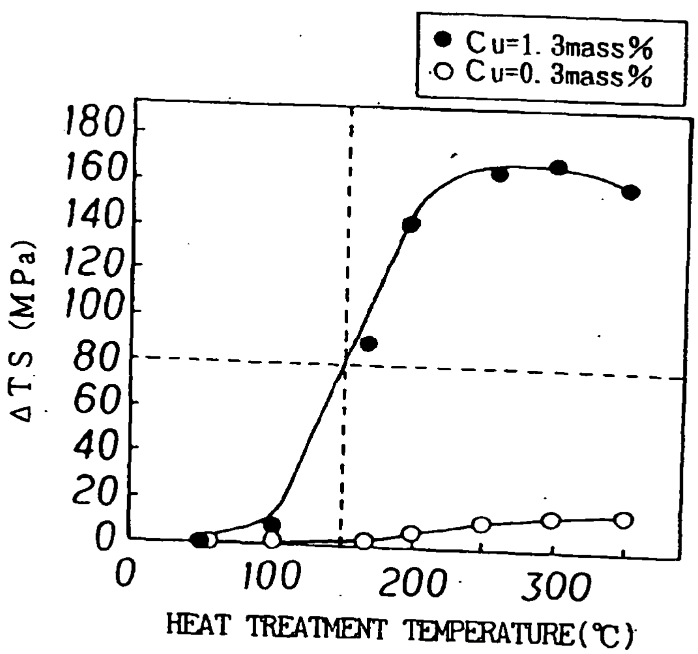

Cu=1. 3mass%
Cu=0. 3mass%
ΔTS (MPa)
180
160
140
120
100
80
60
40
20
0
0
100
200
300
HEAT TREATMENT TEMPERATURE(℃)

Fig. 5

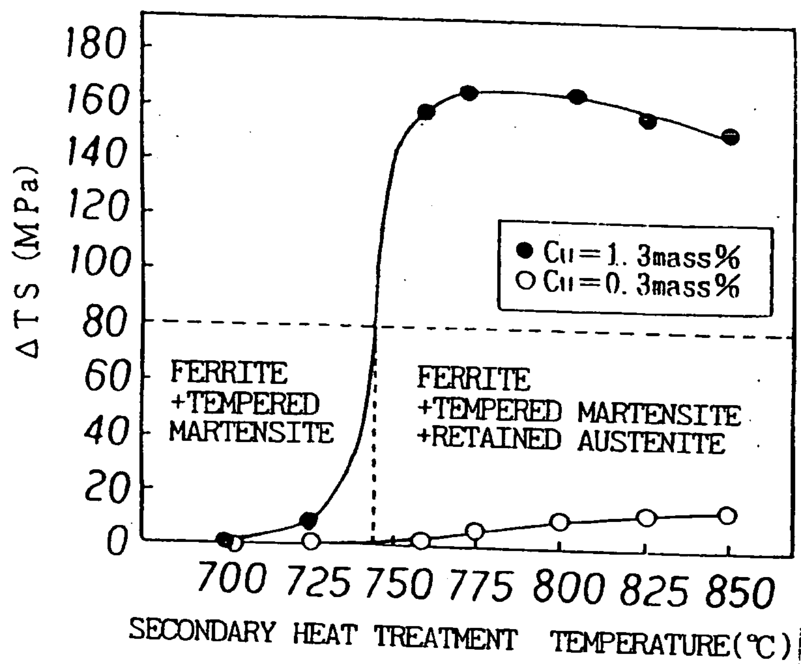
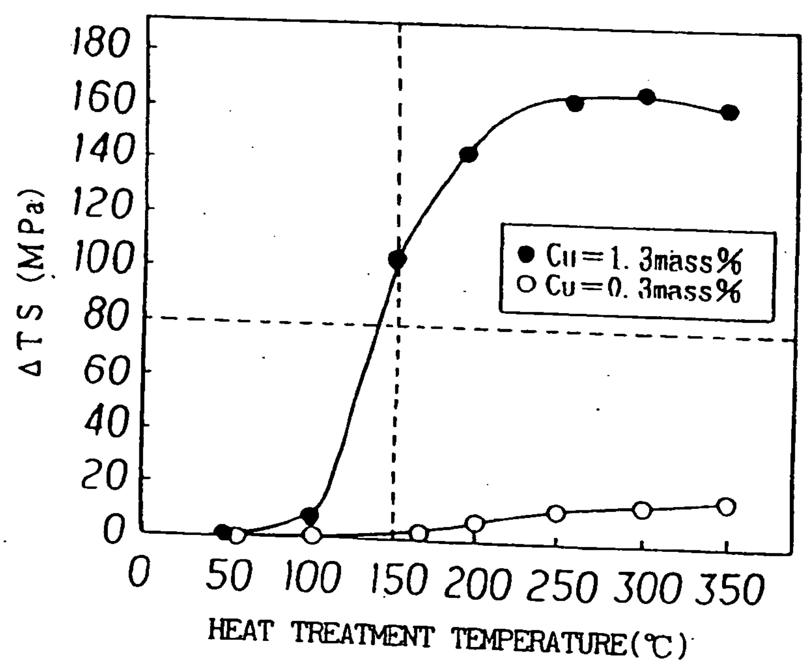

180
160
140
120
100
80
60
40
20
0
ΔTS (MPa)
● Cu=1. 3mass%
○ Cu=0. 3mass%
FERRITE
+TEMPERED
MARTENSITE
FERRITE
+TEMPERED MARTENSITE
+RETAINED AUSTENITE
700 725 750 775 800 825 850
SECONDARY HEAT TREATMENT TEMPERATURE(℃)

Fig. 6


180
160
140
120
100
80
60
40
20
0
ΔTS (MPa)
● Cu=1. 3mass%
○ Cu=0. 3mass%
0 50 100 150 200 250 300 350
HEAT TREATMENT TEMPERATURE(℃)